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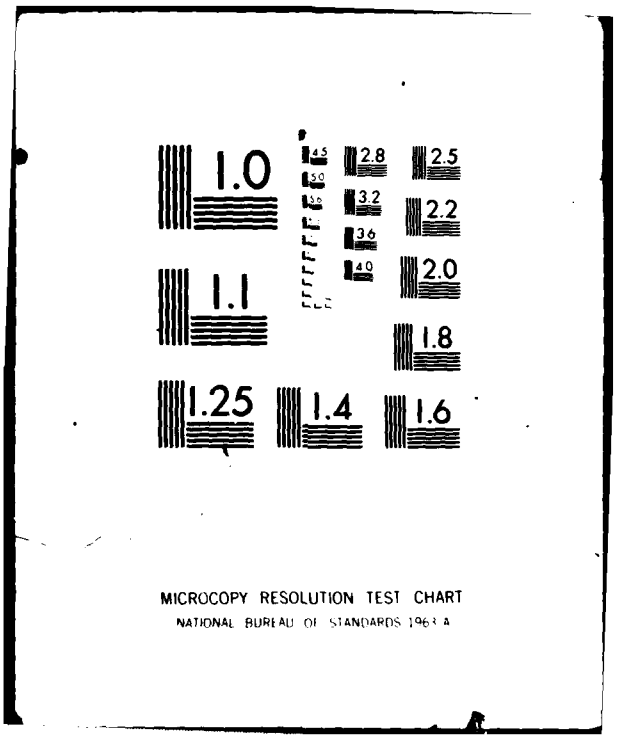
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Microstructural Changes Following Superplastic Deformation of Commercially Pure Titanium

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20. Abstract

Small specimens of commercially pure titanium have been subjected to creep testing in a specially designed furnace at temperatures between 815 and 925°C, followed by rapid cooling to room temperature. The characteristic failure mode was found to be dependent upon temperature and creep stress. at low temperature and/or high stress specimens developed a long diffuse neck whereas at high temperature and/or low stress a short localized neck developed. After testing specimens were examined by optical and electron microscopy. Dynamic recrystallization was observed in some of the former type of test piece. A tentative rationale for the observed behaviour is proposed on the basis of the strain rate sensitivity of the flow stress and the occurrence of grain growth.

Keywords: titanium, creep, strain rate sensitivity, enhanced plasticity.

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This report has been reviewed by the Information Office (EOARD/CI) and is releasable to the National Technical Information Service (NTIS). At NTIS it will be releasable to the general public, including foreign nations.

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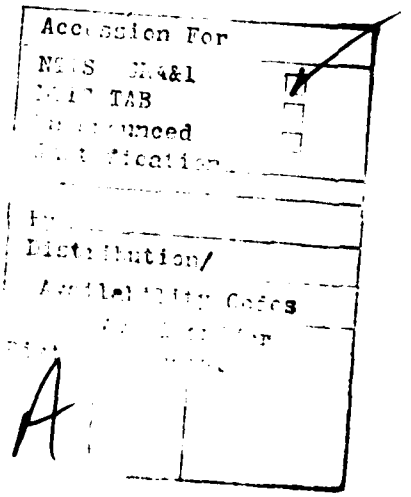

INTRODUCTION

It is well known that certain titanium alloys can be deformed superplastically and this phenomenon has found application in the processes of isothermal forging and blow-forming. The alloy widely subjected to these procedures is probably the $\alpha+\beta$ alloy Ti-6Al-4V and much effort has been directed towards understanding the deformation mechanisms. A complication with this material is that direct observation of the deformed structure is not possible since the $\beta \rightarrow \alpha$ or α' phase change occurs on cooling from the deformation temperature and, consequently, this aspect has not been extensively investigated.

A research programme was therefore initiated to deform very small titanium alloy specimens at high temperature followed by rapid cooling in an attempt to freeze-in the high temperature structure. This paper reports the results of a preliminary evaluation of the equipment using straightforward commercially pure titanium. C.p. titanium was chosen primarily because no phase change was expected and the feasibility of retaining the high temperature structure could therefore be studied. This alloy was also of interest since there is relatively little data available concerning the high temperature deformation of α-Ti alloys.

EXPERIMENTAL

The apparatus consisted of a small creep furnace 25 mm long and 9 mm in internal diameter, containing an insert designed to minimise the temperature gradient. Before the testing programme began the actual temperature gradient was measured over a range of temperatures by attaching thermocouples at several points along the length of a specimen. It was found that, between

800 and 900°C, over the central 4 mm of the gauge length the temperature only varied by ±1.5° and over the central 6 mm by ±6°. The entire apparatus was supported inside an evacuated bell jar (<6 x 10^{-5} mm Hg). Prior to each test, thermocouples were attached to each end of the specimen gauge length and the temperature distribution along the specimen was therefore accurately known. The furnace itself could be raised or lowered to ensure that the gauge length was positioned exactly at the centre of the hot zone.

Specimens 3 mm wide and 6 mm gauge length were prepared from c.p. titanium sheet which had been rolled to 0.3 mm thickness and recrystallized to give a grain size of ~ 10 μm. The specimens were mounted in pairs, side by side and ~ 1 mm apart, but only one was actually loaded and subjected to creep. The other served as a microstructural reference sample so that, by comparison of the two, the structures of strained and unstrained material could be examined after identical times at the same temperature.

After reaching a satisfactory vacuum the furnace was switched on and the desired temperature was attained in about 15 mins. The furnace position was finely adjusted, the predetermined load released and the creep experiment begun. Experiments generally lasted from 6-60 mins., depending on the temperature and initial creep stress, and the specimens were rapidly cooled after fracture by admission of cold argon to the chamber. An L.V.D.T. was used to provide the strain/time record of the creep test.

It should be noted that the experiments were performed at constant load and that the stress therefore changed during the test, at first slowly and linearly while deformation was uniform and then increasingly rapidly when necking began. The parameter chosen for study was, therefore, the initial creep rate, defined as the first linear portion of the ε vs. t curve,

since this could be determined after a few minutes creep and was thus uninfluenced by possible changes in temperature gradient, as the specimen began to move down the hot zone, or the presence of an incipient neck. These two factors can have a significant effect on the later stages of experiments in which high elongations and long times are encountered.

RESULTS

Initial testing was carried out at 815°C but the unexpected results led to two further series of tests at 885 and 925°C. The first corresponds to testing in the single phase alpha field, the second to the two phase region and the third to the single phase beta field and examination of the control specimens confirmed this. Data from the tests are presented in Fig.1 and it is clearly seen that they all fall on three straight lines. Creep is most rapid in the single phase beta condition and lowest in the single phase alpha condition as expected because of the temperature differences

However it was found that the measured elongations after fracture were not as anticipated and the modes of fracture were markedly different. Specimens tested in the single phase alpha field yielded elongations of 120-140% and they failed after the development of a long diffuse neck whereas specimens tested in the beta field failed after much smaller elongations, typically 40-80%, and showed the normal type of localized necking. The behaviour of specimens tested in the alpha+beta field was a little more complex and the data can be roughly divided into two regions about the point X, Fig.1. At low stress and initial creep rate the elongations were comparable to those of the beta specimens whilst at high stress levels elongations of up to 150 were obtained and the mode of fracture resembled the long diffuse neck of the single phase alpha specimens. The characteristic forms of the profiles after testing are shown in Fig.2.

Optical metallography showed that the structure of specimens tested in the alpha field varied from equiaxed at the grip, Fig. 3(a), to heavily elongated near the fracture surface, whilst the volume immediately around the fracture had recrystallized and appeared strain-free, Fig.3(b). Recrystallization nuclei could also be discerned in the heavily deformed region adjacent to the completely recrystallized tip. In the control specimens there was no apparent change in grain size between the grip and the gauge length centre following the period at high temperature.

Tests in the $\alpha+\beta$ field showed recrystallization close to the tip in material subjected to high initial creep rates but not at lower initial creep rates. The structure of the latter type of specimen is shown in Fig. 4 (a) & (b). Specimens from the beta field merely exhibited a small degree of elongation of the prior beta grains near the final fracture surface but pronounced grain growth had occurred.

A limited amount of transmission electron microscopy was performed on material from close to the fracture surface. Specimens crept in the alpha field developed a clear subgrain structure, Fig. 5(a), and recrystallization nuclei were frequently found, Fig.5(b). High creep rate tests on $\alpha+\beta$ material yielded similar structures but in lower creep rate tests recrystallization nuclei were not found and there was less tendency to subgrain formation, although subgrain boundary migration did still occur into regions of high dislocation density, Fig.5(c). A further new feature was the appearance of heavily jogged dislocations containing helical segments, Fig.5(d). As expected the $\beta \rightarrow \alpha$ transformation had removed all reliable evidence of the deformation process in the β-field.

DISCUSSION

The high ductility of the alloy deformed in the alpha field and the sharp decline above the beta transus or at lower stress levels were completely unexpected. A consideration of this behaviour must also take into account that the different ductility levels are characterised by different fracture profiles. The high elongations and long diffuse neck, comprising the entire gauge length, strongly suggested that superplasticity was involved. Optical metallography showed this to be incorrect since one of the features of superplasticity is the absence of any large scale change in microstructure[1] and elongated grains were found even in the two phase condition where superplastic behaviour would be most likely[2].

Instead it is proposed that, in the alpha specimens, the strain rate sensitivity, (S.R.S.) of the flow stress is such that incipient necks are stabilized by an increased amount of work hardening in that region. The strain rate sensitivity index is ~ 0.3 from the experimental data but this reflects only the initial stages of deformation and even larger values may apply at a later stage in the test. Nevertheless a value of 0.3 would tend to give such behaviour. The long diffuse neck develops because incipient necking occurs rather more frequently at the gauge length centre than at the ends due to grip constraint and the small temperature gradient. These tests might thus be termed slow strain rate tests rather than conventional creep tests.

In the beta specimens rapid grain growth occurs due both to the higher temperatures and to the high self diffusivity of β-Ti. As a consequence of this larger grain size the S.R.S. of the flow stress is lower and is unable to stabilize the incipient nack and normal localized fracture follows.

The behaviour of α+β specimens at higher applied loads is similar to that noted for alpha specimens but as the load is reduced it is probable that

the scope of the log τ vs. log $\dot{\varepsilon}$ curve decreases[3] although the variation of the initial strain rate with stress cannot reflect such a change. Once again, strain rate hardening is no longer rapid enouch to stabilize the necks.

It is believed that the occurrence of recrystallization is coincidental in these events and that it occurs only at a late stage in some of the tests after a critical level of strain has been reached. This critical level cannot be reached at lower loads since normal recovery processes, such as dislocation climb and subgrain boundary migration, serve to prevent the accumulation of strain. Damage accumulation is more rapid at higher loads and the necessary critical level can be reached at some stage near the end of the test.

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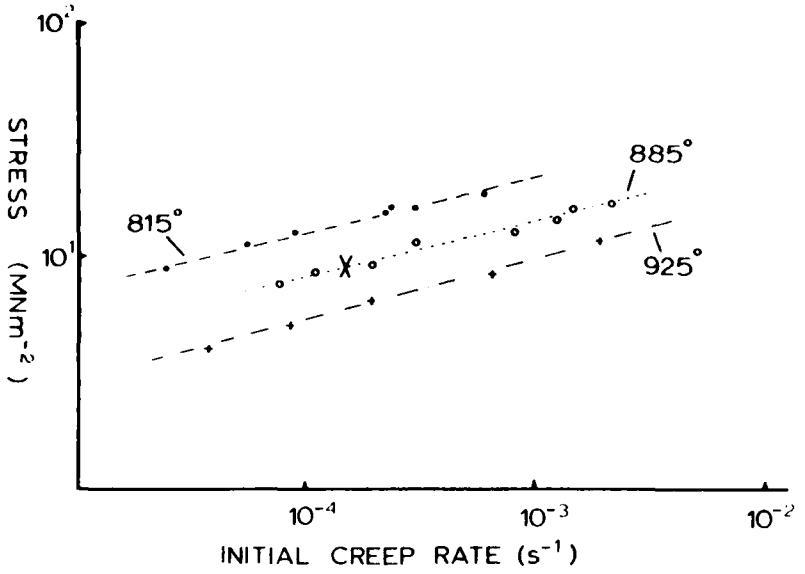


Fig.1. Creep test data (log σ_{app} vs. log $\dot{\varepsilon}$).

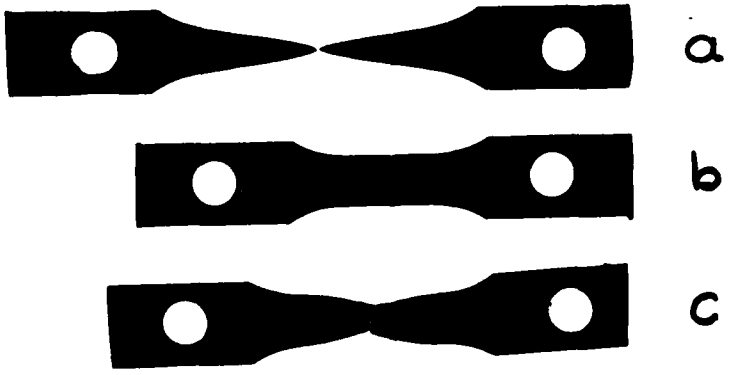
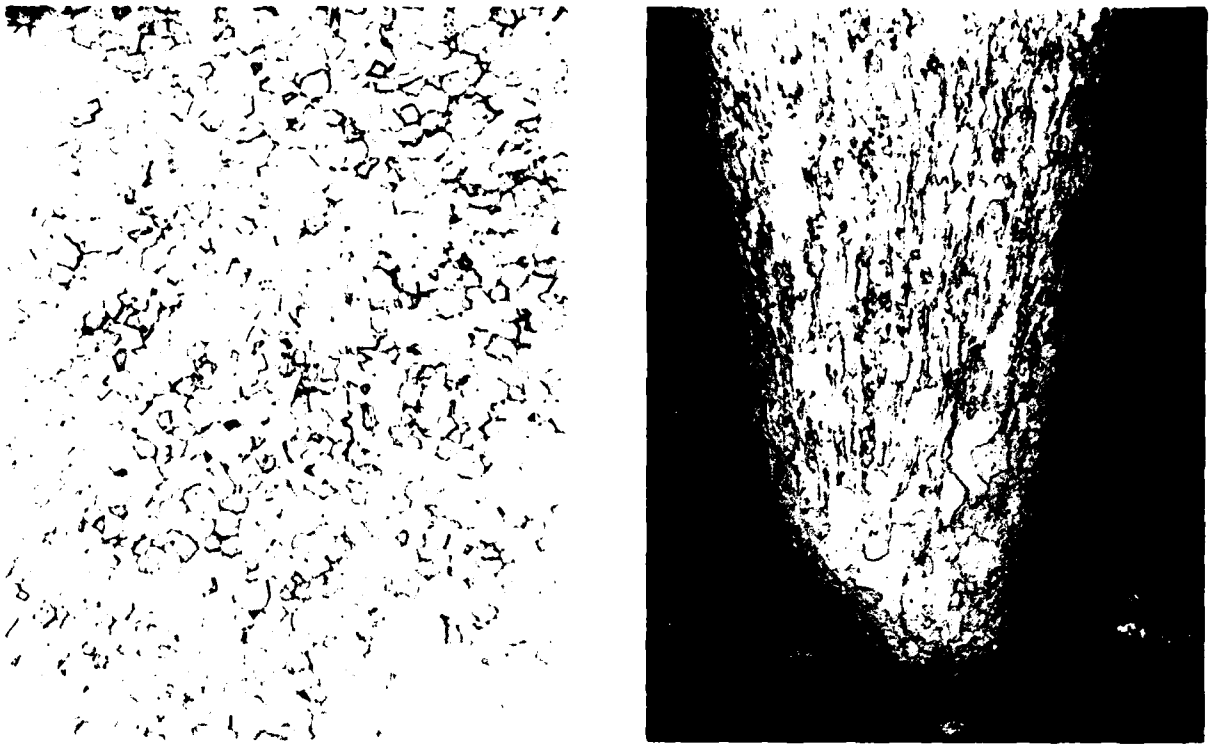


Fig.2. Characteristic profiles: a) crept at 815°C, b) uncrept, c) crept at 925°C.



a b

Fig.3. a) Grip section of specimen tested at 815°C, b) crept at 815°C showing recrystallization adjacent to tip. Magnification x 200.

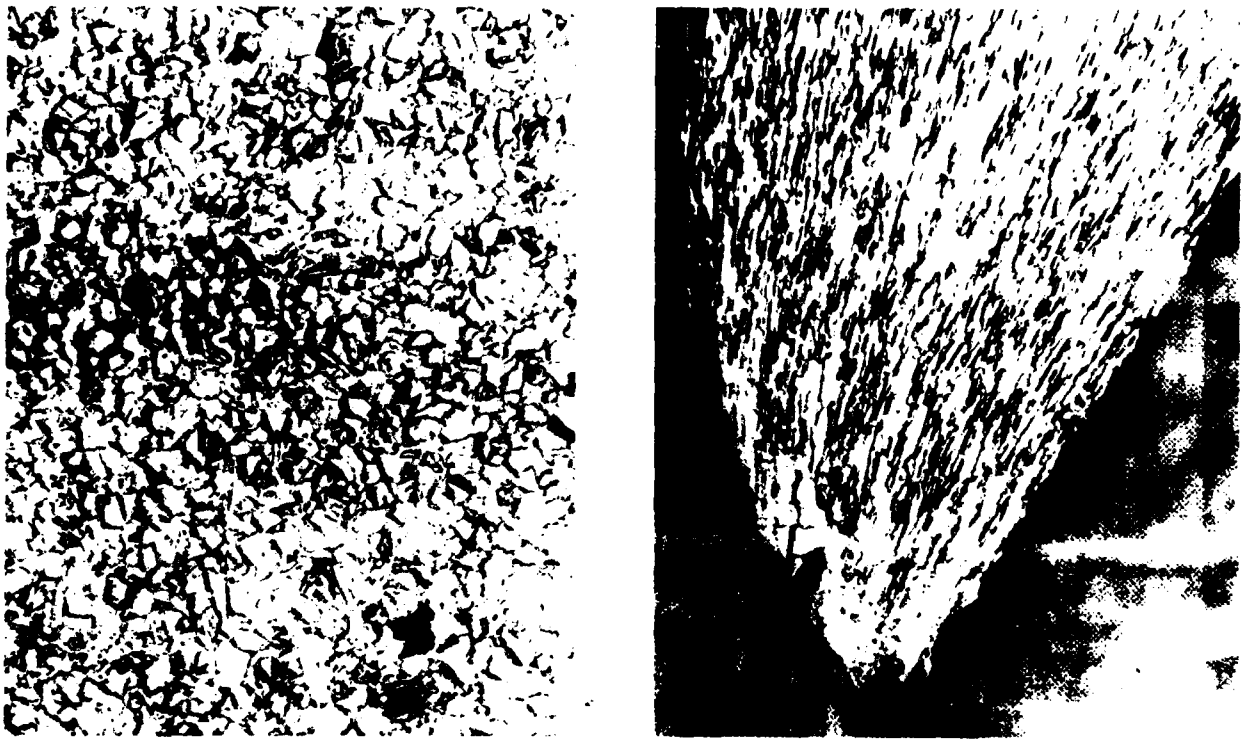


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Fig.4. a) Grip section of specimen tested at 885°C, b) crept at [illegible]°C. Magnification x 200.

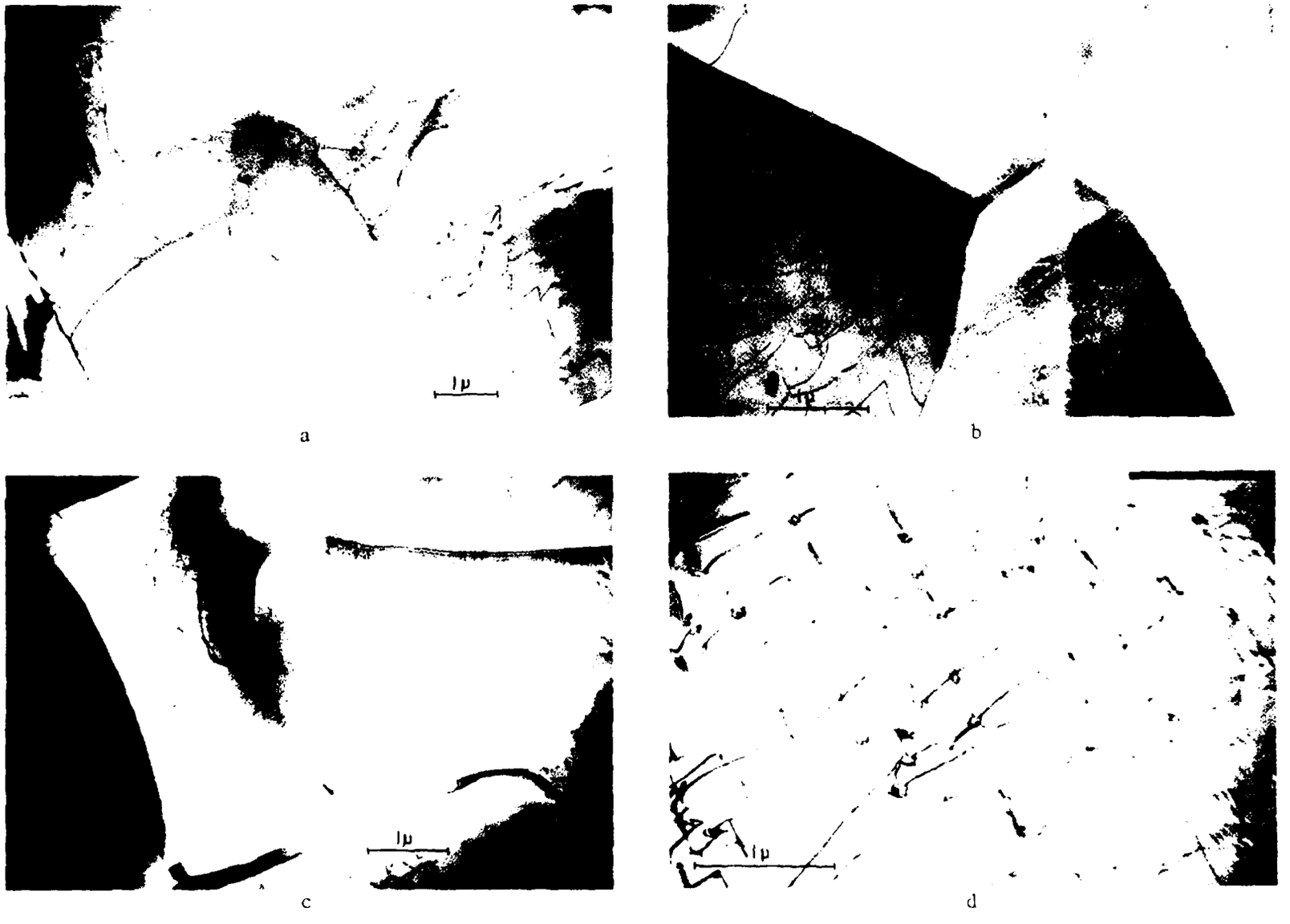
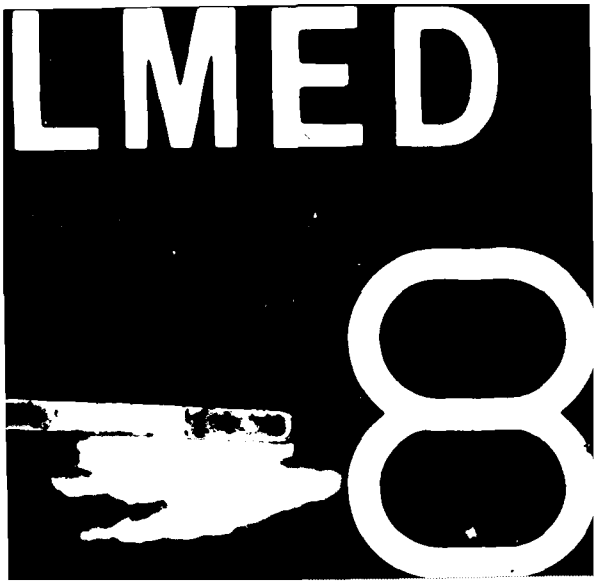


Fig.5. a) Crept at 815° showing subgrain formation,
b) showing recrystallization nucleus,
c) crept at 995° showing subgrain boundary migration,
d) jogged dislocations with helical segments.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